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The effect of high strain rate deformation on intermetallic reaction during ultrasonic welding aluminium to magnesium

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ABSTRACT

High power ultrasonic spot welding (USW) is a low heat input solid-state joining process that may offer a solution for welding difficult dissimilar-material couples, like magnesium (Mg) to aluminium (Al) for automotive body applications. However, the high strain rate dynamic deformation in USW has been claimed to accelerate inter-diffusion rates in dissimilar joints. The interfacial reaction between Al, AA6111, and Mg AZ31 alloys has been studied as a function of welding energy. For the optimum welding condition of 600 J (0.4 s) the reaction layer thickness was already $\sim$5 μm thick. Intermetallic reaction centres were found to nucleate within microwelds at the interface at very short welding times and spread and grow rapidly to form a continuous layer, composed of two sub-layers of $Al_{12}Mg_{17}$ and Al_3Mg_2. Interface liquation was also found for longer welding times at temperatures below the recognised lowest eutectic reaction temperature in the Al–Mg binary system. Modelling has been used to show that the solid state reaction kinetics were over twice the rate expected from parabolic growth predictions made using rate constants obtained under static test conditions. The reasons for this discrepancy and the depressed melting reaction are discussed.

1. Introduction

Aluminium (Al) and magnesium (Mg) are increasingly being substituted for steel in the automotive industry to reduce weight and to increase fuel efficiency [1]. It is also becoming widely recognised that the future of automotive production will be based on a multi-materials approach, allowing more efficient use to be made of the best attributes of different classes of materials [2–4]. This will inevitably result in a greater need to join dissimilar combinations of metals, including aluminium and magnesium alloys. Unfortunately, fusion processes are very difficult to apply successfully to the welding of Al to Mg alloys because the welds are embrittled by the rapid formation of intermetallic compounds (IMC), owing to the high rates of diffusivity in the liquid phase [5,6]. As a result, solid state techniques like friction stir welding (FSW) and friction stir spot welding (FSSW) are attracting increasing interest for dissimilar joining applications (e.g. [7–9]). To date, it has been shown that defect-free Al to Mg welds can be achieved by FSW [9,10], but a significant level of intermetallic compound formation is still observed, which reduces the mechanical properties of the joints [5,7–9,11]. In FSW care must also be taken to control the temperature in order to avoid liquation caused by the low melting point eutectic reactions present in the Al–Mg system (see Fig. 1) [12].

The intermetallic compounds β-Al_3Mg_2 and γ-$Al_{12}Mg_{17}$ have been consistently reported to form in the nugget zone of Al to Mg FSWs [7,8,11,12]. These two phases have also been identified in a recent TEM study by Firouzdor and Kou [7]. To date, other possible intermetallic phases present in the Al–Mg system (e.g. R or ε, and λ [13,14]) have not been observed in welding. However, in FSW the reaction products are broken up into fine particles and redistributed by the severe deformation near the tool, making analysis of their growth behaviour and reaction kinetics difficult [15]. Studies of static diffusion couples between Al and Mg have shown that, when undeformed, the IMC reaction layer typically develops with two continuous sub-layers: $Mg_{17}Al_{12}$ on the Mg side and Al_3Mg_2 on the Al side of the interface [16–18]. The growth kinetics of the overall IMC layer and sub-layers follow the expected parabolic behaviour. However, important in the context of dissimilar welding, the growth rate is abnormally rapid due to the high rate of diffusion through the reaction layer and, in particular, in the β-Al_3Mg_2 phase, which results in a low activation energy [17,19].

Ultrasonic spot welding (USW) is an alternative solid state joining technique to FSSW that can produce welds with a lower energy input [20] and is therefore of interest for joining reactive dissimilar metals, like aluminium and magnesium. Although USW

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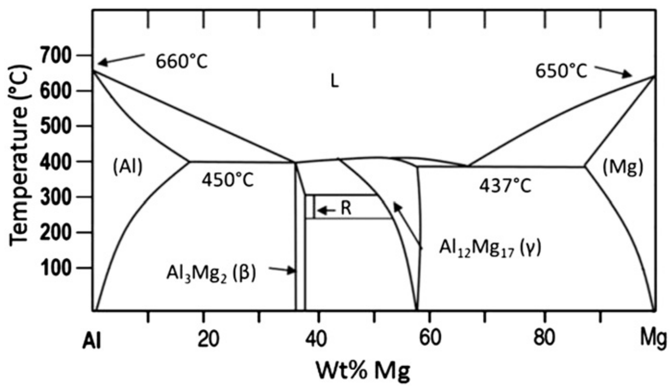

Fig. 1. The binary Al–Mg phase diagram adapted from [7] with the eutectic reaction temperatures indicated.

has been used since 1950s to successfully join thin foils with various dissimilar material combinations [21], higher power welding systems have only been applied to thicker (1–2 mm) automotive sheet relatively recently (e.g. [20,22,23]). Research on lower power USW shows that bonding occurs at moderate temperatures (< 300 °C) and is dominated by contact mechanics, with any deformation localised to the weld faying surfaces [21,23,24]. Weld formation initially involves ultrasonic vibration breaking down the surface oxide layer between contacting asperities, resulting in local adhesion and the formation of microwelds [20,22]. It has been found that when welding thicker (~1 mm) automotive sheet, with higher power systems, optimum weld properties are only obtained when the microbonds fully coalesce across the weld interface. In similar material Al welds this has been observed to require plastic deformation to propagate throughout the entire sheet thickness between the sonotrode tips and in comparison involves considerably greater welding energies [22]. For example, welding energies of the order of 700 J are required to produce optimised welds in 1 mm thick Al sheet that exhibit the desired nugget pull-out failure behaviour [20,22]. Because of the higher energy input, when ultrasonic spot welding 1 mm thick sheet, Chen et al. have measured peak temperatures at the interface in Al welds to be above 400 °C [23].

As well as the more substantial temperature rise seen in high power USW of Al sheet, an important consideration in dissimilar joining is the high strain rate ($\varepsilon^{\bullet} \sim 10^3\ s^{-1}$ [25]) dynamic deformation caused by the USW process. It has been claimed that this can result in accelerated reaction kinetics. For example, in their study of Zn to Al dissimilar joints, Gundaz et al. [25] have reported both enhanced inter-diffusion rates and a significantly depressed melting point, which they attributed to the presence of the large concentration of deformation-induced vacancies generated by USW. Chen et al. have also noted an effect of deformation induced vacancies on post-weld natural ageing in Al USWs [23]. Potentially, such effects could have an important influence on the intermetallic reaction kinetics when ultrasonic welding Al to Mg.

Currently, little work has been published on the role of deformation on the intermetallic reaction behaviour seen when ultrasonic welding Al to Mg, especially with the higher power systems required for thicker gauge automotive sheet. Previous work by Panteli et al. has sought to optimise the welding parameters for Al to Mg using the USW technique. This work showed that pre-grinding the surfaces of the weld coupons and increasing the clamping force (up to 1.9 kN) improved strength. However, even under optimised conditions, they could not produce welds of sufficient strength to fail by nugget pull-out and this was attributed to the rapid formation of a brittle intermetallic layer at the joint interface [26].

Here, we have investigated in detail how the intermetallic reaction layer forms and influences the joint performance when performing dissimilar ultrasonic spot welds between the aluminium and magnesium automotive alloys AA6111 and AZ31, in 1 mm thick gauge sheet. The evolution of the IMC reaction layer and growth rate has been studied as a function of welding time/energy and related to accurate measurements of the interface temperature. To determine the effect of the high strain rate dynamic deformation inherent to the USW process, the reaction layer thickness has been compared to predictions of the growth rate under static conditions, as well as to thermodynamic calculations of the melting point at the joint interface. It is also worth noting that the conclusions of this analysis are applicable to other joining process involving intense deformation, such as in FSW, where the interface behaviour is more difficult to interpret owing to the more extensive material flow that can occur in the weld zone.

2. Experimental

The welds investigated were produced between aluminium AA6111-T4 and magnesium AZ31-H24 alloy, 1 mm thick, sheets. The hot rolled Mg sheet was first prepared by grinding off the thick oxide using 320 grit SiC paper, whereas the cleaner Al sheet was welded in the as-rolled condition. Both materials were thoroughly degreased with ethanol prior to welding. Ultrasonic spot welding was performed with a Sonobond dual-head spot welder operating at a frequency of 20.5 kHz and a nominal power of 2.5 kW (machine setting). The materials were welded between two 9 mm × 6 mm sonotrode tips, which had nine parallel ridged teeth orientated perpendicular to the direction of vibration (see Ref. [22] for full details). Laser vibrometry was used to measure the amplitude of oscillation at the sontrode tip, which was ~5–6 μm. Welding was performed with a constant power setting, *P*, for increasing welding times, *t*, and weld energy, *U* (where $U \sim P \times t$) using a clamping force of 1.9 kN. The actual power delivered to the weld members varied in the range 1200–1500 W depending on the coupling between the sheers and welding tips. The welds were made at the centre of 25 mm overlap on 100 mm × 25 mm coupons held with light manual clamping. The weld temperatures were measured as close as possible to the join line at the weld centre, the hottest location in the weld [23], using embedded sacrificial 0.5 mm K-type thermocouples. The thermocouples were inserted through a groove in the top aluminium sheet. Temperature measurements were repeated several times and only the results that gave 'reliable' heating and cooling curves were retained (i.e. time–temperature histories with a similar profile to that expected from modelling with no inflections). DSC analysis was also performed on pre-welded samples, using a heating rate of 10 K s^{-1}, to measure the incipient melting point at the interface under static conditions.

Tensile lap shear testing was carried out on the welded samples using a cross-head speed of 0.5 mm min^{-1}, with both the peak load and fracture energy (area under the load–displacement curve) measured. Metallographic samples were prepared from the weld cross sections using standard procedures, with oil-based lubricants, and finished using water-free colloidal silica suspension. Microhardness tests were carried out on flat, polished samples using a Vickers CSM micro-indentation hardness testing. Imaging was carried out by conventional optical microscopy and with a Philips XL30 or FEI Sirion FEG SEM. Phase identification of the reaction layer was performed by high resolution EBSD with a step size of 0.01 μm on samples prepared by removing the surface layer across the weld interface with a FEI Quanta 3D dual beam FIB, to obtain a strain free surface.